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## **Ultra-Low Carbon Bainitic Steels for Heavy Plate Applications**

Final Report  
Submitted to  
David Taylor Research Center, U.S. Navy

by  
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ULTRA-LOW CARBON BAINITIC STEELS  
FOR HEAVY PLATE APPLICATIONS

FINAL REPORT

SUBMITTED TO

DAVID TAYLOR RESEARCH CENTER, U.S. NAVY

BY

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ULTRA-LOW CARBON BAINITIC STEELS  
FOR HEAVY PLATE APPLICATIONS

ABSTRACT

A new family of ultra-low carbon bainitic (ULCB) steels for heavy plate applications has been investigated. These new steels have been developed as possible candidates to replace traditionally heat treated HTS and HY steels. One major attraction of the ULCB steels is that they do not need to be heat treated to obtain their final mechanical properties. The ULCB steels obtain their high strength and high resistance to both ductile and brittle fracture through the proper combination of alloy design and thermomechanical treatment.

## ADMINISTRATIVE INFORMATION

This report discusses work performed as part of the HSLA-100 Steel Materials Block Program. The technical point of contact for this task was I. Caplan (DTRC Code 0115). Work was performed at this Center under work unit number 1-2814-950 in FY-89. The work was conducted under the supervision of M. Vassilaros, Fatigue and Fracture Branch, DTRC Code 2814.

## INTRODUCTION

A series of modified ultra-low carbon bainitic (ULCB) steels were investigated as possible candidates to substitute for conventional quenched and tempered (Q+T) high yield strength (HY) steels. One of the major attractions of ULCB steels over Q+T steels is that, ULCB steels are capable to attain an excellent combination of mechanical properties in the as-hot rolled condition in sections up to 50 mm (2 in.) thick, without the need of additional heat treatment. Another major advantage of ULCB steels over Q+T steels is the weldability behavior. Because it is well known that both the overall weldability and weldment toughness are inversely related to carbon equivalent values, especially at high carbon contents, the weldability of the Q+T high yield strength steels is relatively poor. Since ULCB steels can develop high strength and toughness with low carbon content, these steels should exhibit good weldability and HAZ toughness.

This report summarizes the final results of the research program sponsored by the David Taylor Research Center of the U.S. Navy at the University of Pittsburgh in the development of a new family of ULCB plate steels which must not only exhibit the same high level of strength and toughness attainable in HY steels, but also, they must be much more easily weldable. The three-year study of the feasibility of developing ULCB steels for use by the U.S. Navy at the 80 and 100 Ksi (550 and 690 MPa) yield strength levels has recently

been completed. The major findings of this study are as follows:

1. Mechanical properties equal to or better than those of HY-80 and HY-100 can be obtained in ULCB steels in the as-rolled condition.
2. The strength of ULCB steels is controlled by the bainite-start ( $B_s$ ) transformation temperature. The  $B_s$  temperature is a direct function of the composition of the steel, and may also be influenced by the thermomechanical processing.
3. The resistance to brittle and ductile fracture of the ULCB steels is directly controlled by the metallurgical condition of the austenite after hot rolling.
4. The optimum hot rolling conditions are those which induce grain refinement of austenite through repeated recrystallization during high temperature hot rolling followed by intense "pancaking" of these small grains by heavy reductions taking place below the 100% recrystallization stop temperature of the austenite.

#### **BACKGROUND**

During the last 20 years, efforts have been undertaken to provide alternatives to the conventional Q+T HY type of steels. Prior to 1980, there was only one type of plate steel



which was capable of achieving both high strength and good low temperature toughness in moderate to heavy sections. These were the martensitic steels which were Q+T to the desired yield strength level, generally in the range of 80 to 100 Ksi (560 to 690 MPa). When these steels are heat treated to a yield strength of 80 and 100 Ksi, they are called HY-80 and HY-100, respectively. The nominal chemical composition of these steels is 0.15% C - 2.5% Ni - 1.5% Cr - 0.3% Mo. Experience has shown that the base plate properties of both HY-80 and HY-100 steels are more than adequate for most applications.

The properties which are the most critical in plate applications are strength, toughness and weldability. While the base plate properties of HY-100 in a properly heat treated (Q+T) condition are usually more than acceptable, the weldability of these steels is troublesome and the weldment properties are sometimes poor. Perhaps the main reason for the poor weldability of steels such as HY-100 is that they are the product of an alloy design philosophy (ca 1890) which existed before fusion welding even began to be considered as a means of joining steel plates. According to Lancaster, the oxy - acetylene welding of plates became an accepted practice in 1916, and arc welding with fusible electrodes at the time of World War II <sup>(1)</sup>. Hence, multi-pass submerged arc welding (SAW) did not become a technological reality until about a half-

century after the heat treatable low alloy base plates were originally configured. Based on these circumstances, it is perhaps not surprising that the traditional Q+T steels are less than ideally weldable.

One of the major factors governing the weldability of a plate steel is the immunity of the HAZ to cold, delayed or underbead cracking which may occur subsequent to the fusion process. This problem was examined in some detail by Graville, and some of his results are shown in Figure 1 <sup>(2)</sup>. Graville indicates that the susceptibility of the HAZ of a given steel to cold cracking depends on both the composition of the steel and the manner in which it has been welded. For example, when the carbon content and carbon equivalent value place the steel in region II, the susceptibility will depend on the details of the welding process, e.g. heat input, thickness and preheat. Steels falling in region III are susceptible under all welding conditions, while steels of region I are immune. Hence, according to Figure 1, the problem of cold cracking of the HAZ can be avoided in plate steels of sufficiently low carbon content.

A second major problem concerning weldability is the low temperature toughness of the HAZ after multi-pass SAW. Recent work has revealed that both the composition and microstructure of the HAZ control its low temperature toughness<sup>(3)</sup>. The important features appear to be the presence of a coarse-

grained region immediately adjacent to the fusion line and the presence of martensite - austenite (M-A) islands within this coarse - grained zone. The influence of these M-A islands on the fracture toughness of a coarse - grained HAZ in linepipe steel is shown in Figure 2. Since the amount and size of these island would be expected to vary with the carbon content of the steel, steels with lower carbon levels would be expected to exhibit lower amounts of the M-A islands, and, hence, the higher low temperature HAZ toughness.

From the results shown in Figures 1 and 2, it seems apparent that any new approach to the alloy design of highly weldable plate steel should be centered on having the new steel contain the lowest possible carbon content. Unfortunately, however, virtually all steel strengthening mechanisms rely, to some degree, on carbon content. The fundamental question then becomes one of how to achieve sufficient strength in the base plate at carbon levels which are low enough to ensure both easy weldability and good weldment properties.

#### HISTORY OF ULCB STEELS

There are essentially three types of steel microstructures capable of achieving high strength levels: ferrite - pearlite, bainitic, and martensitic. A comparison of the approximate strength levels attainable in low carbon, low

alloy steels with each of these structures is shown in Figure 3<sup>(4)</sup>. Since the required tensile strength is approximately 950 MPa (138 Ksi), it is clear that this level of strength can only be reached by the bainitic or martensitic structures. What remains to be explored is whether either of these structures is compatible with the requirement that the base plate have a very low carbon content ( $C \leq 0.04\%$ ) for good welding performance.

Martensitic steels are low carbon (0.1 - 0.4%C), low alloy steels which exhibit sufficient hardenability to be through-hardened under appropriate cooling conditions. The hardenability of these steels is dependent not only on the substitutional solutes such as Mn, Mo, Cr, Ni, etc., but also on the carbon content itself. It appears that carbon content not only has a powerful and direct influence on hardenability<sup>(5)</sup>, but also has a strong influence on the contribution to hardenability made by substitutional elements<sup>(6)</sup>. Hence, as the carbon content approaches zero, the ability to form martensite in these low alloy steels is diminished appreciably. Furthermore, as the carbon content approaches zero, the difference in strength between martensite and bainite also approaches zero<sup>(7)</sup>.

While it appears to be difficult to obtain martensite in ultra-low carbon, low alloy steels of reasonable thickness, it is possible to form bainite in these steels. The CCT diagrams

of steels typical of the HY grades indicate that the nose of the proeutectoid ferrite/pearlite reactions is located at a cooling rate of approximately  $0.25^{\circ}\text{C/s}$  and the nose of the bainite reaction is located at a cooling rate of approximately  $20^{\circ}\text{C/s}$ <sup>(6)</sup>. Hence, it is easy to form martensite in HY - type of steels by water quenching. However, at very low carbon contents, the bainite nose moves to the left and essentially overlays the  $M_s$  line. Therefore, it is possible to obtain bainite over a wide range of cooling rates in these very low carbon, low alloy steels. As early as in 1967, the potential benefits of a very low carbon bainitic steels were being explored. These early results showed that yield strengths of 750 MPa (110 Ksi) with good notch toughness could readily be achieved in an as-rolled 0.03%C - 0.7%Mn - 3.0%Ni - 3.0%Mn and 0.07%Nb bainitic steel<sup>(9)</sup>. In this work, McEvily et al. also recognized the need to avoid or at least minimize second phase particles such as carbides and nonmetallic inclusions which could act as nucleation sites for either voids, leading to premature ductile fracture, or cleavage cracks, leading to premature brittle fracture. Although McEvily et al. did not explicitly discuss weldability, their steel would appear to offer good welding behavior, based on Figures 1 and 2. Since the earlier work of McEvily et al. other low-carbon bainitic steels have been developed and evaluated. These included the Mn-Mo-Nb<sup>(10-12)</sup> and the 4%Mn FAMA steels<sup>(13)</sup>. The work by Coldren

et al. showed that the two major factors responsible of the poor low temperature toughness in their plate steels were grain boundary  $\text{Fe}_3\text{C}$  and high carbon M-A islands. In the late 70's Nippon Steel Corporation developed an ultra-low carbon bainitic steel for large diameter, high pressure linepipe for Artic applications<sup>(14)</sup>. The typical composition of this steel was 0.02%C - 2.0%Mn - 0.4%Ni - 0.3%Mo - 0.04%Nb - 0.02%Ti and 0.001%B. This steel was controlled rolled and exhibited a yield strength in the range 480 to 550 MPa (70 to 80 Ksi) in up to 20 mm (0.8 in) thick plates with good low temperature toughness. Both the weldability and HAZ toughness of this steel were very good.

In summary, technical and economical forces have resulted in the need for new plate steels which exhibit not only adequate strength and toughness in the hot rolled condition but also good weldability and weldment properties. Thus what follows describes the results of the research program conducted on the physical metallurgy of ULCB plate steels intended for moderate -to- heavy section applications. Of particular interest in this study was the influence of factors such as composition, austenite conditioning (thermomechanical processing, TMP), and post-rolling heat treatments on the final microstructure and mechanical properties of ULCB steels.

### EXPERIMENTAL PROCEDURE

The chemical composition (in wt%) of selected steels evaluated in this investigation is shown in Table I. The steels were either vacuum melted (Nos. 1, 2 and 3) or air induction melted (Nos. 4, 5, 6 and 7). The ingot weight and size was 225 Kg (500 lbs) and 200 mm x 200 mm x 675 mm (8 in x 8 in x 27 in), respectively.

### MATERIALS PROCESSING

To evaluate the influence of reheating temperature and thermomechanical processing on the microstructural conditioning of the austenite, steels 1 through 4 were subjected to the thermomechanical treatment illustrated in Figures 4a and b. For all the steels used in this study, the amount of deformation during roughing and final rolling was maintained constant such that the final plate thickness was always the same, 25.4 mm (1 in). The other steels (5, 6 and 7) were processed using optimum controlled-rolling practices such as the one illustrated in Figure 4c.

The effect of tempering treatments on the mechanical properties of the as-hot rolled ULCB steels was also investigated. The steel samples were tempered for one hour in the temperature range of 500 - 675°C (1022 - 1247°F) and then air cooled to room temperature.

### MECHANICAL PROPERTIES

The mechanical properties of all the steels in both the as-hot rolled and tempered conditions were determined using standard procedures. The tensile properties were determined in sub-size flat tensile specimens with a 25.4 mm (1 in.) gage length. The specimens were cut transverse to the rolling direction. In addition, impact properties were obtained from full size Charpy V-notch specimens. The notch of the specimens was perpendicular to the rolling plane (T - L orientation).

### MICROSTRUCTURE

The microstructural characteristics of both the prior austenite and the final bainitic ferrite were evaluated using standard metallographic procedures involving optical (OM) as well as scanning and transmission electron microscopy (SEM and TEM). The morphology of the controlled rolled austenite grains and the cleanliness of the steels were studied using OM and SEM whereas the fine details of the bainitic structures were examined using SEM and TEM. The metallurgical condition of the prior austenite was assessed through the use of the parameter  $S_v^{(15)}$ . This parameter measures the total area per unit volume of near-planar crystalline defects such as grain boundaries, deformation bands and twin boundaries. The non-metallic inclusions present in the steels were evaluated using the stereological parameter  $\Sigma b$  or total projected length<sup>(16)</sup>. This



parameter was determined using a computer - controlled Bioquant System IV image analyzer.

### **TRANSFORMATION TEMPERATURE**

The transformation start or  $B_s$  temperature was determined for each steel during the period of air cooling from the final rolling pass. The  $B_s$  temperature was taken to be the temperature of the thermal arrest found on the cooling curves. This temperature corresponds to approximately 15% transformation. The average rate of cooling was found to be approximately  $0.1^\circ\text{C/s}$ .

### **MECHANICAL PROPERTIES**

The transverse mechanical properties of the steels in both the as - hot rolled and tempered conditions were determined using standard procedures. The tensile properties were determined in sub-size flat tensile specimens with a 25.4 mm (1 in) gage length. Impact properties were obtained from full - size Charpy V - notch specimens with the notch perpendicular to the rolling plane (T - L orientation). The fracture toughness of the steels according to ASTM - 813 - 87 standards was determined at room temperature. The J - integral method incorporating double - cantiliver beam specimens was applied.

## **RESULTS**

### **TRANSFORMATION TEMPERATURE**

The bainite - start or  $B_s$  temperature measured after various amounts of controlled rolling is shown in Table II.

### **MICROSTRUCTURE**

The microstructure of the prior - austenite, as indicated by the parameter  $S_v$ , revealed the effectiveness of the various TMP practices used during the conversion of the ingots to final plate. The values of  $S_v$  determined from several rolled plates and its influence on the impact energy of the steels at two testing temperatures (-18 and -85°C) is shown in Figure 5. Examples of the microstructure of the as-rolled austenite are shown in Figure 6.

The microstructure of the bainitic ferrite is also revealed in optical micrographs of Figure 7. This bainitic ferrite appears to conform to the so-called granular description<sup>(7)</sup>. The fine internal structure of the bainitic ferrite, as revealed by TEM of thin foils, is shown in Figure 8 for steel 4. The salient features of this structure may be broadly described as follows: (A) the major structural component is a series of elongated laths with the lath boundaries consisting of walls of dislocations: (B) While the laths have an overall general appearance similar to the subgrains and cells which form during recovery<sup>(6)</sup>, the laths

differ in that the laths themselves contain large numbers of dislocations; (C) The axes of the elongated laths and elongated austenite grains are coincident. While several packets of laths of similar spacial orientation appear to form from coarse - grained, recrystallized austenite<sup>(6,7)</sup>, only one set of laths seems to form from pancaked austenite grains; (D) Superimposed on the lath structure is an additional array of dislocations. The scale of this array is much larger than that of the laths, perhaps approaching the size of the parent austenite grains; (E) While some retained austenite and M-A constituent was found, the amount of these phases was quite small and their occurrence rather rare; (F) The only precipitates found in the hot rolled condition were NbCN. The size, distribution and location of these precipitates indicated that they had been formed in the austenite during controlled rolling similar to those often observed in ferrite - pearlite steels<sup>(17-19)</sup>. Virtually no evidence of inter-lath NbCN,  $\epsilon$ -carbide or  $\text{Fe}_3\text{C}$  was found in the as-rolled condition.

The cleanliness or non - metallic inclusion size distribution of the various steels was evaluated using the stereological parameter  $\Sigma b$  or total projected length. The values of  $\Sigma b$  for the various steels are shown in Table III.

#### STRENGTH OF ULCB STEELS

Previous work has shown that composition has a

significant effect on the strength of bainitic steels<sup>(6,10)</sup>. This effect has been related to the influence of composition on the  $B_s$  temperature. An example of this effect is shown in Figure 9 for the steels used in the current study. All tensile data for the ULCB steels are presented in Table IV.

### TOUGHNESS OF ULCB STEELS

The resistance of the ULCB steels to both ductile and brittle fracture is shown in Table III and Figure 10. In this regard, the upper shelf energy (USE) of the standard Charpy V-notch impact energy curve is taken as a measure of the resistance to ductile fracture, while the 50% shear FATT is taken as a measure of the resistance to brittle fracture. The fracture toughness ( $J_{1c}$ ) of a different but similar set of ULCB steels is shown in Figure 11.

## DISCUSSION

### MICROSTRUCTURE

The first level of microstructural interest in this study was the metallurgical condition of the prior - austenite, since good austenite conditioning is a direct result of proper TMP, each element of TMP must be controlled. Hence, the reheating, hot rolling, and cooling stages of TMP may each contribute to the success or failure in obtaining good

austenite conditioning.

It is well known<sup>(15,19,20)</sup>, that the metallurgical state of the austenite which results from hot rolling can be described by the parameter  $S_v$  or effective austenite interfacial area. In this regard, the higher the  $S_v$ , the more effective has been the austenite conditioning during processing. Previous work by Kozasu et al<sup>(15)</sup>, Speich et al<sup>(17)</sup>, and DeArdo<sup>(20)</sup> has led to a better understanding of  $S_v$  and has shown that there are three factors which contribute to  $S_v$ : (1) initial grain boundary area per unit volume of equiaxed grains prior to pancaking, (2) increase in initial grain boundary area caused by change in grain shape during pancaking, and (3) formation of intracrystalline defects during rolling. The first two effects have combined<sup>(17)</sup> in the following expression:

$$S_v = 1/D(1 + 1/R + R)$$

where  $D$  is the austenite grain diameter prior to pancaking, and  $R$  is the reduction ratio. The third effect has been approximated as follows:

$$S_v(\text{ICB}) = 0.63\% (\text{red.} - 30)$$

where  $S_v(\text{ICB})$  is the contribution to  $S_v$  resulting from intracrystalline defects such as deformation bands and incoherent twin boundaries, and % red. is the rolling reduction (%) below the recrystallization - stop temperature<sup>(21)</sup>.

The use of conventional TMP technology to ULCB steels

requires certain modifications. This condition arises because the chemical composition of the steels used in the present study are more highly alloyed than the typical ferrite-pearlite microalloyed steels. In addition, the plates of the current study were 25 mm thick whereas typical ferrite - pearlite linepipe wall thickness is often in the range 12 - 18 mm. The above factors demanded a modification of the traditional view of TMP, because the high levels of substitutional solute in the steels of the present investigation resulted in: (a) changes in the relations governing the solubility of NbCN in austenite, and (b) abnormally high levels of solute drag which hinder the motion of crystalline defects. Furthermore, the thicker plate used in the present study makes TMP more problematic because it is more difficult to achieve homogeneous structures in plates which, during rolling, have gradients in temperature and strain rate in the thickness direction.

The response to TMP of austenite microalloyed (MA) with Nb, V and/or Ti depends importantly upon (a) how the reheat temperature compares to the grain coarsening temperature  $T_{gc}$ , (b) how the temperature of rolling compares to the recrystallization stop temperature  $T_{rx}$ , and (c) how the cooling rate after rolling compares to the CCT diagram. These relationships have been discussed in some detail elsewhere<sup>(19,20,22,23)</sup>. The three critical temperatures of

austenite, i.e.  $T_{GC}$ ,  $T_{RX}$ , and either the  $A_{r3}$  or the  $B_s$ , are all strongly influenced by the type and disposition of the MA elements in the steel. In the case of the present study, the high levels of solute present in the steels altered, as expected, the critical temperatures usually associated with the Nb levels used. It has been shown that elements such as Mn and Mo can increase the solubility of NbCN in austenite. This would then act to lower all three critical temperatures of the austenite. Proper TMP for optimum austenite conditioning must accomodate these changes in critical temperatures. Properly conditioned austenite benefits ULCB plate steels by increasing both the YS and resistance to brittle fracture.

#### **TRANSFORMATION BEHAVIOR**

When low alloy steels of the type investigated in this work undergo transformation during continuous cooling, the resulting microstructure is often a mixture of bainite, retained austenite, carbides, and martensite. An example of the type of CCT diagram for one of the steels used in this investigation is presented in Figure 12. The precise microconstituents present, of course, depend upon the cooling rate. Much can be learned about the bainite in a given alloy system by measuring the  $B_s$  and also the extent to which the  $B_s$  temperature varies with changes in experimental conditions. Since the scale of the transformation products decreases with

lower transformation temperatures, lower  $B_s$  temperatures are associated with higher dislocation densities and smaller bainitic lath sizes. In cases where the transformation temperature is sufficiently high to permit simultaneous adequate atom mobility and solute supersaturation, lower  $B_s$  temperatures are also associated with finer particle dispersions. Hence, several studies have shown an inverse linear relationship to exist between strength and  $B_s$  temperature, as shown in Figure 9.

It is well-known that the addition of appropriate solutes to low carbon steels can: (a) introduce a "bay" of unstable austenite which acts to separate the ferrite "nose" from the bainite/martensite regions on TTT or CCT diagrams, see Figure 12, (b) cause a delay in the formation of proeutectoid ferrite, and (c) depress the  $B_s$  temperature<sup>(24,25)</sup>. Hence, the major factor which controls the  $B_s$  temperature is the solute content of the austenite phase in the steel at the time of transformation. The influence of composition on  $B_s$  temperature has been determined<sup>(29)</sup>.

$$B_s(^{\circ}\text{C}) = 830 - 270C - 90\text{Mn} - 37\text{Ni} - 70\text{Cr} - 83\text{Mo}$$

Although the above equation is very useful to predict the expected  $B_s$  temperature of low to high carbon bainitic steels, it cannot, however, be used to calculate accurately the  $B_s$  temperature of modern ultra-low carbon bainitic steels. From the results of this study the following equation has been



found to predict the  $B_s$  temperature within  $\pm 10.0^\circ\text{C}$  of accuracy for the range of chemical compositions used in this investigation:

$$B_s(^{\circ}\text{C}) = -25.3[10.2 + 1625B + 68.1(C+N) + 46.3(Ti+Nb) + 4.8Mo + 3.3Mn + 2.6Cr + 0.3Ni] + 1173.35$$

Another factor which has a significant influence on  $B_s$  temperature is the recrystallization state of the austenite prior to transformation. The influence of amount of deformation below the recrystallization - stop temperature of austenite on the measured  $B_s$  temperature is shown in Table II. the influence of the retained deformation of austenite has opposite effects depending on the level of alloying in the steel. When the alloy level is relatively low, i.e.  $B_s$  high, the deformation of austenite causes the  $B_s$  temperature to increase. On the other hand, the opposite trend is found in the steel having the higher concentration of alloying and lower  $B_s$  temperature. Effects similar to these have been observed before. It is well-known, for example, that the  $A_{r1}$  or transformation temperature for proeutectoid ferrite increases with the extent of controlled rolling prior to transformation<sup>(21)</sup>. This has been interpreted to be caused by an increase in the strain - induced precipitation of NbCN with strain and the subsequent lowering of the Nb in solution in the austenite at transformation. This lowering of solute Nb then is responsible for higher  $B_s$  temperatures. The mechanism

whereby deformation of austenite lowers the  $B_s$  temperature is less clear. It could be hypothesized that the deformed, richly alloyed austenite might undergo some form of austenite stabilization as is observed in martensitic reactions<sup>(27)</sup>. in this regard, perhaps the high strength of austenite resulting from retained strain, precipitation, and solid solution strengthening would render the austenite less prone to transformation by shear-related mechanisms.

#### MICROSTRUCTURE OF BAINITIC FERRITE

The optical micrographs shown in Figure 7 indicate that the bainitic ferrite found in the experimental steels can be described as granular bainite<sup>(8)</sup>. It should be noted, however, that surface relief studies have shown that the bainitic structure may be far more acicular or needle-like in nature than can be discerned by simple optical microscopy<sup>(9)</sup>. Observations of the fine internal structure of the bainite revealed a lath structure with high dislocation densities in the lath walls, Figure 8. Superimposed on this lath structure was a second dislocation network of a scale much larger than the typical lath dimensions. In general, the microstructure shown in Figure 8 is similar to those found in the literature for fairly comparable compositions and cooling rates<sup>(9,10)</sup>.

Since the bulk carbon content is initially very low in the experimental steels, and with the loss of some carbon to

NbCN during controlled rolling, it would appear unlikely that the solubility limit of carbon in ferrite would be exceeded during transformation. Hence, little if any evidence of  $\epsilon$  or  $\text{Fe}_3\text{C}$  was found in the microstructure. Furthermore, little if any evidence of M-A-C constituent was found.

### STRENGTH OF ULCB STEELS

The strength of hot rolled steels has been often expressed as an expanded Hall-Petch equation<sup>(28)</sup>. Although equations of this type were originally developed to rationalize the strength of ferrite-pearlite steels, they can also be helpful in understanding the strength of ULCB steels. A typical example is:

$$YS = YS_0 + \Delta YS_s + \Delta YS_p + \Delta YS_t + \Delta YS_d + K_y D_f^{-1/2}$$

where  $YS$  is the observed yield strength,  $YS_0$  is the lattice friction stress and  $\Delta YS_s$ ,  $\Delta YS_p$ ,  $\Delta YS_t$  and  $\Delta YS_d$  are strengthening increments caused by solid solution, precipitation, texture and dislocation effects, respectively. The last term,  $K_y D_f^{-1/2}$ , is the contribution to strength by ferrite grain size. The major contributions to strength in ULCB steels appear to be  $\Delta YS_s$  and  $\Delta YS_d$ , since there is virtually no coarse or fine second phases, and no formal grain size. There are three types of dislocations present in the ULCB microstructure: (a) in the bainitic lath boundaries, (b) in a uniform distribution along the laths, and (c) in a network of a scale larger than the

lath size. All three dislocation distributions will contribute to strength. While the first two densities most likely result from the shear-type nature of the transformation, itself, the third distribution may be inherited from the unrecrystallized austenite. Hence, as the alloy level increases, the lath size decreases and the overall dislocation density increases. This results in higher strengths with lower  $B_s$  temperatures. The third dislocation density is weakly dependent of both composition and microstructural state of the austenite. The results of this investigation have shown that controlled rolling contributes approximately 100 to 120 MPa to the yield strength of ULCB steels<sup>(29)</sup>. The relationship between the yield strength, composition and the equivalent austenite grain diameter for the ULCB steels investigated in this program is shown as follows:

$$YS \text{ (MPa)} = 25[10.2 + 1625B + 68.1(C+N) + 46.3(Ti+Nb) + 4.8Mo + 3.3Mn + 2.6Cr + 0.3Ni] + 116d^{-1/2}$$

where the amount of the alloying elements is in wt% and  $d$  is the equivalent austenite grain diameter,  $(4A/\pi)^{1/2}$ .

### TOUGHNESS OF ULCB STEELS

In the traditional view of toughness, the resistance to ductile fracture is controlled primarily by the non-metallic inclusions while the resistance to brittle fracture is

controlled by some appropriate grain size. It is well-recognized, for example, that the ductile fracture characteristics in the transverse direction are related to the amount, size and shape of the non-metallic inclusions present in the steel<sup>(30,31)</sup>. This trend is also found in the present study as shown in Table III and Figure 10. A comparison of steel cleanliness ( $\Sigma b$ ) with the upper shelf energy for the steels in Table II of comparable strength level, e.g. steels 1, 3, 4, and 7, clearly illustrates the expected inverse relationship.

The factors that control the resistance to brittle fracture were somewhat more complex, as indicated in Table III and Figure 10. It was found that the resistance to brittle fracture was controlled by three factors: (a) the metallurgical condition of the parent austenite, (b) the cleanliness of the steel, and (c) the strength level. The influences of these three factors are shown most clearly in Figure 10, where the fraction of ductile fracture found on the fracture surfaces of the Charpy impact specimens is plotted as a function of test temperature. These observations in terms of the fracture appearance transition temperature, FATT, are presented in the following equation:

$$\text{FATT}(\text{°C}) = -191 + 0.14\text{YS}(\text{MPa}) + 0.83\text{D}(\mu\text{m}) + 0.04\Sigma b(\mu\text{m}/\text{mm}^2)$$

where YS is the yield strength, D is the effective austenite grain size and  $\Sigma b$  is the total projected length of the non-

metallic inclusions. The information from the above equation indicates that the effective grain size of the austenite is the most important factor, followed by the strength level and the cleanliness in order of importance in increasing resistance to cleavage fracture.

These three factors also exerted a strong influence on the fracture toughness determined at room temperature using the J-integral method, Figure 11. The highest fracture toughness was found for the steel with a high  $S_v$  and the lowest  $\Sigma b$  (curve 3). The lowest toughness was found for the steel with the highest  $\Sigma b$  (curve 4). The load-displacement curve with multiple rapid load drops (curve 1) is indicative of the occurrence of cleavage fracture. Hence, cleavage fracture can occur at room temperature in ULCB steels in this highly restrained test when the steels exhibit both poor austenite conditioning and poor steel cleanliness.

Previous studies have shown that hard particles can have an adverse effect on brittle fracture resistance because the particles can act as nuclei for cleavage cracks<sup>(32)</sup>. In general, these hard particles could be inclusions, carbides and/or high carbon martensite islands<sup>(10,32)</sup>. Hence, any reduction in the amount of these particles would be translated into a lower amount of cleavage fracture, a lower FATT and a higher fracture energy. Clearly, the cleanliness of the steel is controlled by the steelmaking and solidification practice

used in producing the steel. The presence or absence of large carbides (i.e.  $\text{Fe}_3\text{C}$ ) and high carbon martensite islands, on the other hand, are related to the bulk carbon content and the overall alloy design. The low carbon content of the steels of the present study largely preclude the formation of carbides and martensite islands.

While the benefits of good austenite conditioning have been appreciated for some time in ferrite - pearlite steels<sup>(19-23)</sup>, the same is not true for the bainitic steels. Perhaps the one notable exception was the application of controlled rolling to bainitic steels as described by Irvine in 1969<sup>(33)</sup>. Irvine showed how the impact transition temperature of bainitic steels could be lowered through controlled rolling of the austenite, at the same time increasing the yield strength<sup>(33)</sup>. This is similar to the trend observed in the present study, i.e. controlled rolling improved both strength and resistance to brittle fracture in the ULCB steels.

The results of the present study indicate that the resistance to cleavage fracture of the bainitic ferrite is principally related to the effective grain size of the austenite from which it transformed. While the nucleation of cleavage is controlled by the inclusions, the growth of the cleavage cracks appears to be related to the austenite grain size. Detailed studies of the early stage of cleavage cracking have revealed that neither the polygonal<sup>(34)</sup> nor the bainitic<sup>(29)</sup>

ferritic microstructure offers intrinsic resistance to a growing cleavage crack. In these previous studies of the early stages of cleavage fracture, virtually no evidence of partially cleaved polygonal or bainitic grains was found. The cleavage cracks extended between ferrite grain boundaries in the polygonal ferrite and between prior austenite grain boundaries in the bainitic ferrite. Since the austenite boundaries are the only effective barriers to the growth of cleavage cracks in ULCB steels, the strong influence of  $S_v$  or the effective austenite grain size on resistance to cleavage cracking can be rationalized. One possible scenario is presented in Figure 13. In Figure 13, the length of an initiated cleavage crack is assumed to be  $C'$  and the length at which this crack starts to grow catastrophically under the prevailing conditions is  $C^{*(35)}$ . According to the experimental observations, the initiated crack will slowly grow until it reaches the austenite boundaries at which time its growth will be temporarily arrested. If the arrested crack length is larger than  $C^*$ , then the crack will grow through the boundaries and macrofracture will occur. On the other hand, if the arrested crack length is smaller than  $C^*$ , then the crack will not grow and macrofracture will be avoided. Since the arrested crack length is related to the austenite grain size, the central role played by  $S_v$  in suppressing cleavage fracture can be understood.



### EFFECT OF TEMPERING TREATMENTS

The behavior of ULCB steels to aging treatments is presented in Figure 14. This figure shows that ULCB steels have basically a two stage response to tempering treatments. In the first stage, the steels appear to be fairly insensitive to tempering treatments within a given temperature range. The second stage is manifested by a sharp decrease in hardness. The onset in decrease of hardness appears to occur at a critical temperature which depends on the Mo level of the steel. The overall variation in hardness during tempering treatments suggests that the overall strength of ULCB steels is maintained by a balanced process between precipitation and dislocation recovery. The major contribution of tempering treatments was manifested by increasing the impact toughness without a decrease in the strength of the steels. Figure 15 shows the beneficial effect of tempering on the resistance to both cleavage and ductile fracture of ULCB steels. This behavior can be rationalized in terms of two concomitant and offsetting events: (a) the loss in strength due to dislocation annihilation and rearrangement is most likely balanced by the increase in strength due to precipitation, and (b) the increase in impact toughness is due to subgrain formation and elimination of stresses introduced during the bainitic transformation reaction.

### CONCLUSIONS

In summary, the present work has investigated the alloy design - thermomechanical processing - microstructure - property relation of a series of new ultra-low carbon bainitic steels. The results have shown that proper alloy design and appropriate TMP will lead to the development of ULCB steels with an excellent combination of strength and toughness in the hot rolled condition. The level of austenite conditioning has a dramatic effect on the  $B_s$  temperature and, therefore, also in strength. The two major contributions to strength are solid solution and dislocation hardening. The toughness of the ULCB steels is controlled by the prior austenite grain morphology or  $S_v$ , the steel cleanliness is related to the nucleation of cleavage cracks while the  $S_v$  controls the growth of cleavage cracks. Hence, high resistance to cleavage fracture requires clean steels and proper austenite conditioning.

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**TABLE I. COMPOSITION (wt%) OF EXPERIMENTAL STEELS**

Steels	C	Mn	P	S	Cr	Ni	Mo	Nb	N	Ti	B
1	.021	.94	.005	.004	---	1.41	1.49	.052	.006	.016	.001
2	.017	1.01	.006	.004	---	3.15	3.02	.055	.001	.013	.0011
3	.018	.51	.002	.003	---	3.05	1.53	.053	.006	.020	.002
4	.024	.95	.002	.002	.48	3.54	1.76	.050	.009	---	---
5	.026	.92	.003	.003	.42	3.58	2.59	.053	.008	---	---
6	.027	.95	.008	.002	.45	3.63	3.05	.055	.005	---	---
7	.026	1.01	.007	.002	.49	3.58	1.70	.052	.006	---	---

**TABLE II. INFLUENCE OF AUSTENITE DEFORMATION BELOW RECRYSTALLIZATION  
TEMPERATURE ON Bs TEMPERATURE**

# STEEL	Bs Temperature ( ° C) at Indicated Deformation				
	0%	20%	50%	Bs <sub>(20%)</sub> -Bs <sub>(0%)</sub>	Bs <sub>(50%)</sub> -Bs <sub>(0%)</sub>
1	590	608	614	18	24
4	540	553	552	13	12
5	504	497	493	-7	-11
7	564	562	569	-2	5

TABLE III. COMPARISON OF QUANTITATIVE PARAMETERS OF AUSTENITE GRAIN SIZE AND INCLUSIONS WITH MECHANICAL PROPERTIES OF STEELS TESTED

# Steel	YS MPa(ksi)	Sv mm <sup>-1</sup>	$\Sigma b$ $\mu m/mm^2$	USE J(ft-lbs)	FATT °C(50%)
1	750(109)	69	1054	80(59)	0
2	945(137)	93	432	83(61)	-15
3	690(100)	114	676	111(82)	-50
4	752(109)	179	390	127(94)	-68
5	800(116)	256	1416	81(60)	-55
6	834(121)	209	339	115(85)	-48
7	745(108)	221	243	153(113)	-75



**TABLE IV. TENSILE PROPERTIES OF EXPERIMENTAL STEELS**

# STEEL	YS MPa(ksi)	UTS MPa(ksi)	el %	RA %	CVN, J(ft-lbs)		
					100° C	-18° C	-85° C
1	750(109)	883(128)	18.3	63.2	80(59)	70(52)	16(12)
2	945(137)	1073(156)	15.7	51.0	83(61)	53(39)	13(10)
3	670(97)	807(117)	23.5	66.0	111(82)	84(62)	26(19)
4	752(109)	924(134)	19.2	61.8	127(94)	113(84)	71(53)
5	779(113)	972(141)	15.9	58.2	81(60)	63(47)	35(26)
6	834(121)	1027(149)	16.0	61.3	115(85)	106(79)	69(51)
7	745(108)	917(133)	18.3	67.4	153(113)	146(108)	71(53)

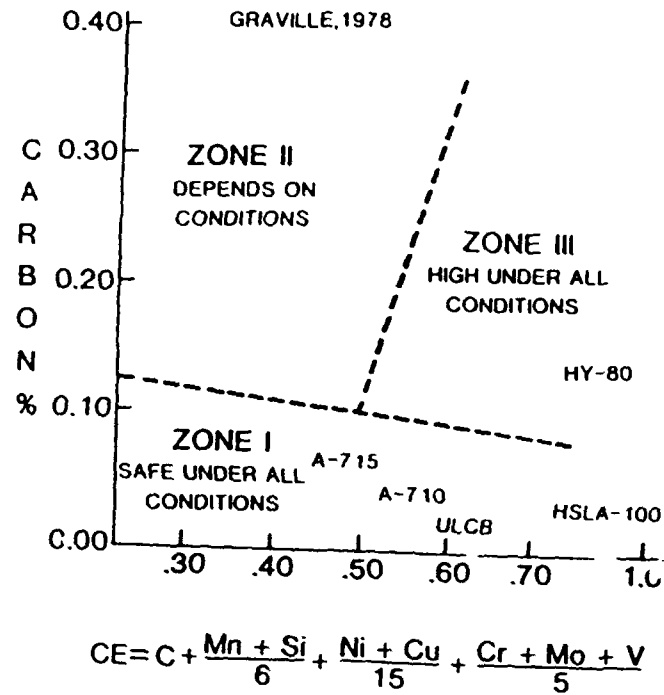


Figure 1. Susceptibility to heat affected zone cracking .

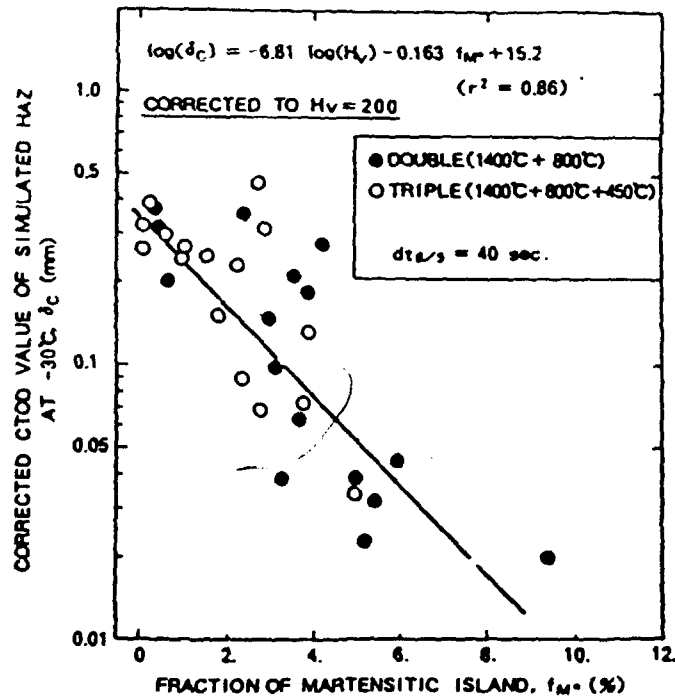


Figure 2. Relation between CTOD value and fraction of martensitic islands of simulated HAZ .

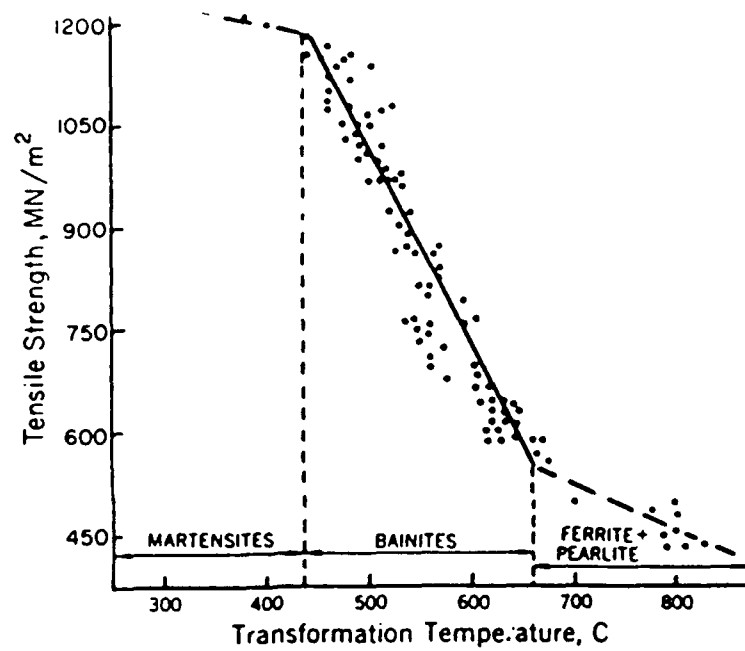


Figure 3. Effect of transformation temperature on the strength of ferrite-pearlite and bainitic structures.

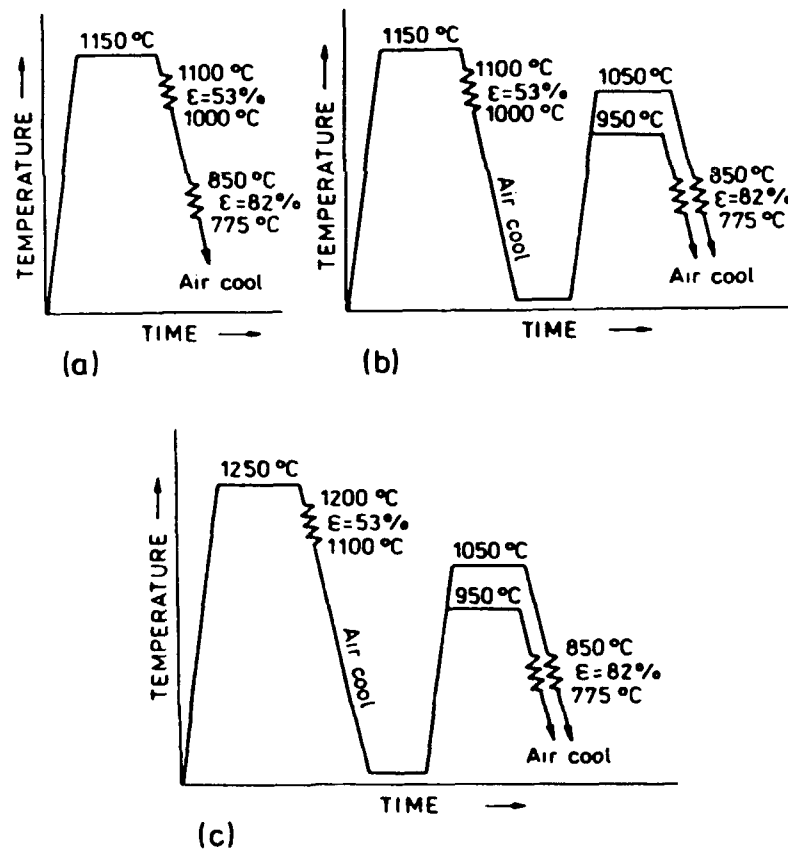


Figure 4. Thermomechanical processing of ULCB steels.

# Effect of Sv on CVN of ULCB-100 Steels

(HR condition, 1 in. plates)  
T-L direction

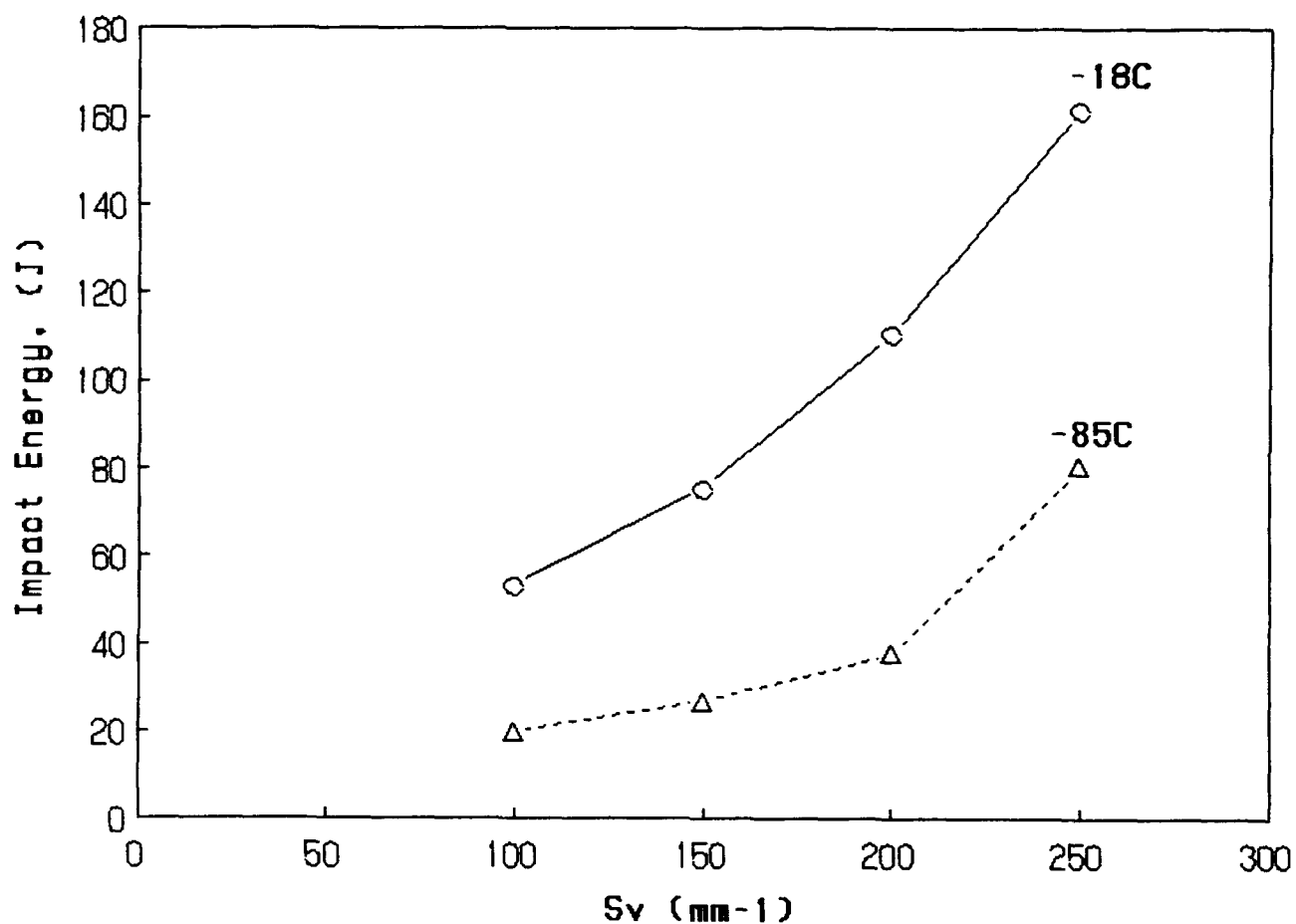


Figure 5. Effect of austenite conditioning on impact toughness



Figure 6. Optical micrograph of controlled rolled austenite (400 x).



Figure 7. Optical micrograph of bainitic ferrite microstructure (1000 x).

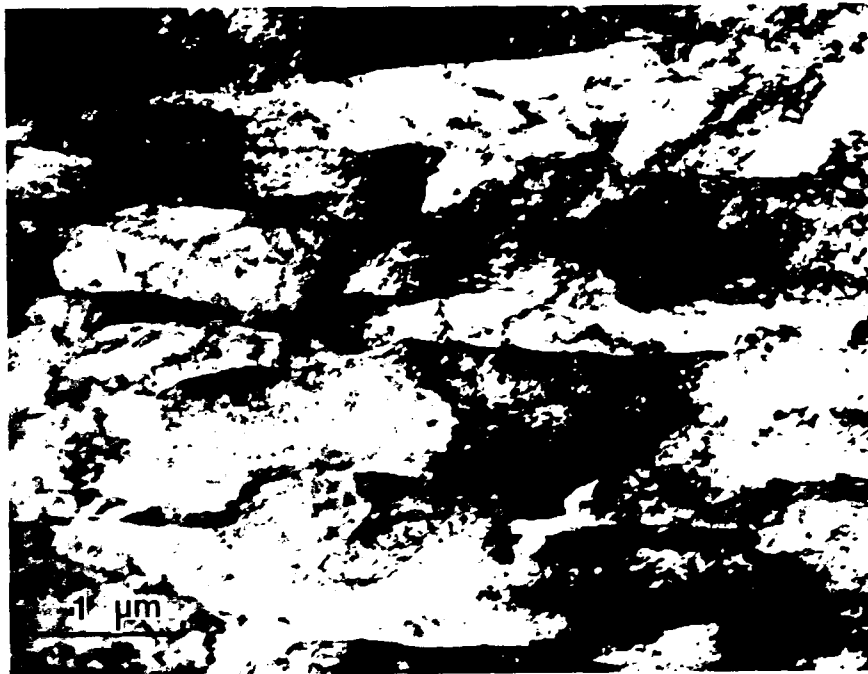


Figure 8. TEM micrograph of controlled rolled ULCB steel revealing the fine internal structure of the bainitic ferrite.

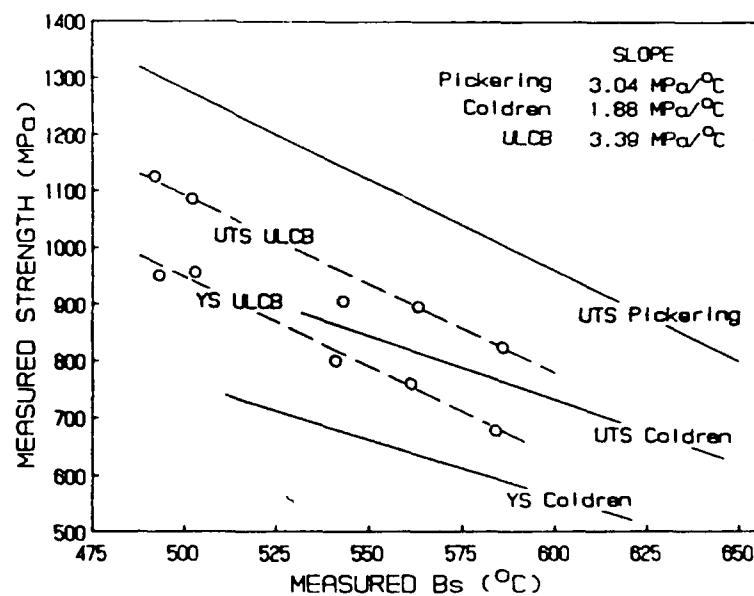


Figure 9. Strength and  $B_s$  temperature relationship for TMP ULCB steels. Also included are data from Pickering<sup>(4)</sup> and Coldren<sup>(10)</sup> studies.

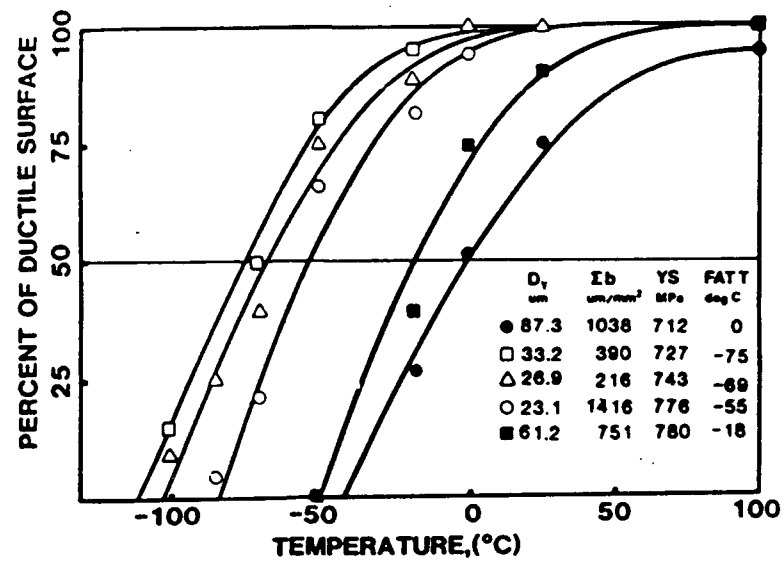


Figure 10. Factors which influence the resistance to cleavage fracture in ULCB steels in Charpy V-notch impact test.

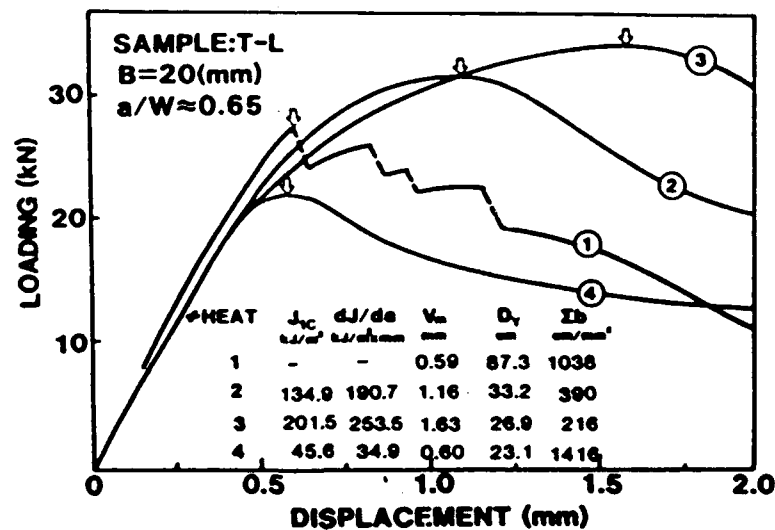


Figure 11. Factors which influence the fracture toughness of ULCB steels. J-integral tests performed at room temperature.

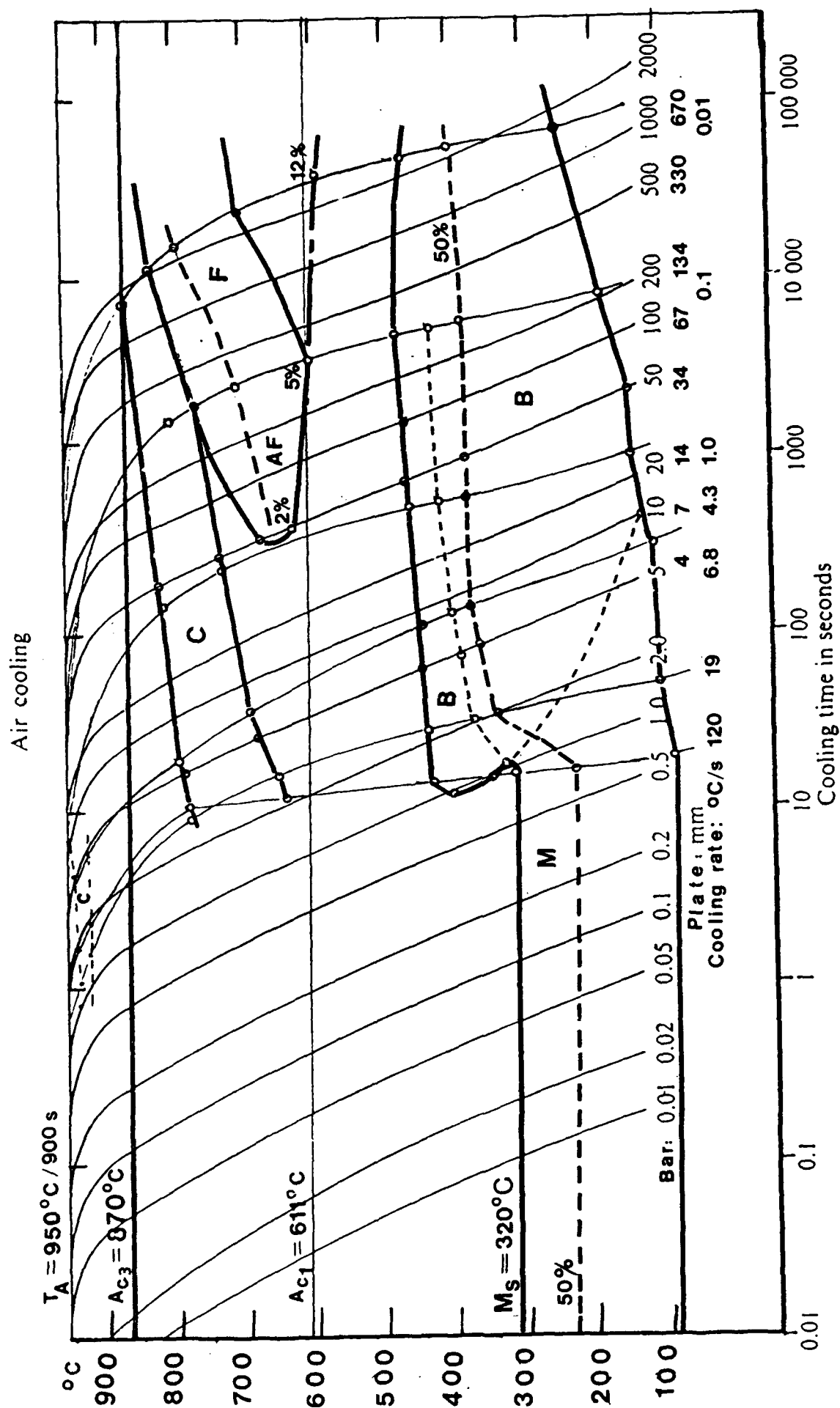


Figure 12. CCT diagram of ULCB steel (No. 7)



## CLEAVAGE FRACTURE AND ROLE OF $S_v$ .

Assume

$C'$  = initiation length  
=  $f(\gamma_{100}, \text{Matrix}, \text{NMI})$

$$C^* = \text{critical length} = \frac{2G\gamma}{\pi(1-\nu)\sigma^2}$$

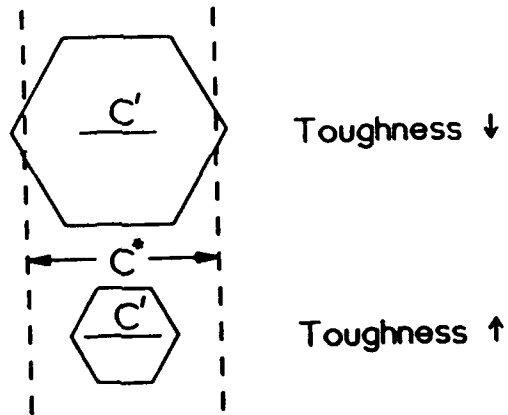


Figure 13. Schematic illustration of role of grain size in suppressing growth of cleavage crack.

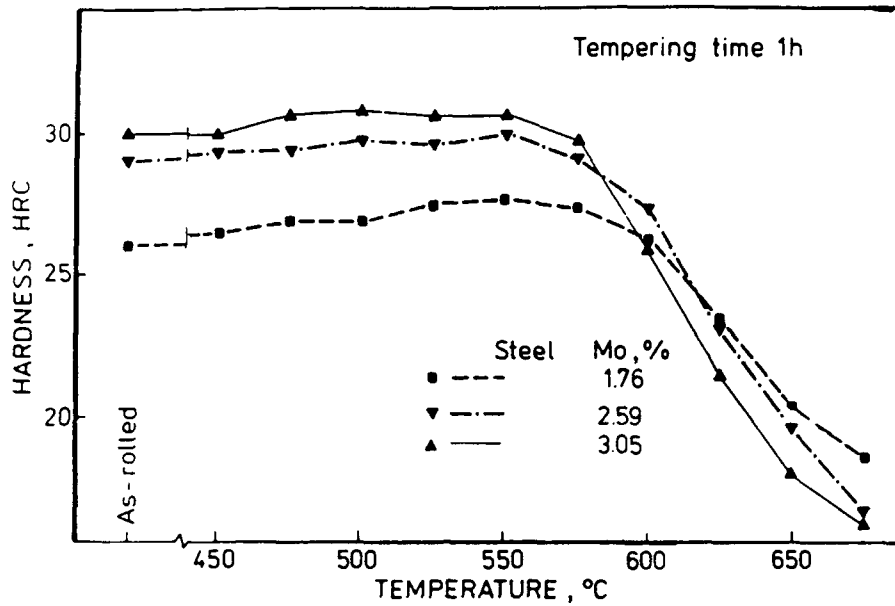


Figure 14. Tempering behavior of ULCB steels.

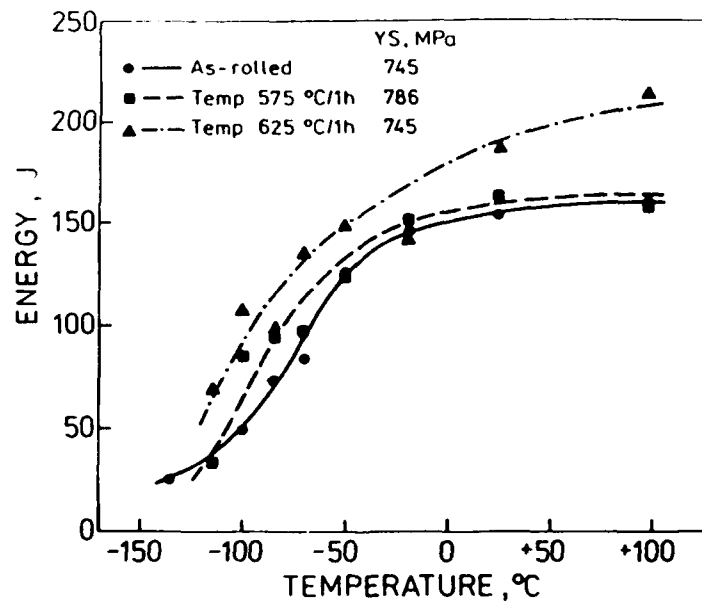


Figure 15. Comparison of impact properties between as-rolled and as-tempered ULCB steels.

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